photographic plates, BF_3 counters, and pulse ionization chambers. They all registered an intensity of less than ½ percent of the intensity with the lithium target in place. A survey of competing processes which might produce results such as appear here was made. Available information on the mass of Be^6 eliminates the possibility of the reaction $Li^6(p, n)Be^6$ for protons of 3.5-Mev energy. The (n, p) reaction cross sections of constituents of the emulsions are all quite low compared with the (n, p) scattering cross section. When this consideration is combined with the fact that the relative abundance of hydrogen is 3.0 times that of any other constituent, the contribution of the reaction processes can be eliminated as a source of error.

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Multiplication Processes for Slow Moving Dislocations

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THE slip bands observed in the plastic deformation of crystals show that on a typical active slip plane there is about 1000 times more slip than would result from the passage of a single dislocation across the plane. One of us[1] has discussed a possible explanation of this in terms of the reflection and multiplication of dislocations which have acquired velocities approaching that of sound. However, there is as yet no available experimental evidence for fast dislocations and a recent theoretical estimate[2] of the energy dissipated by a moving dislocation indicates that under typical conditions the terminal velocity of a dislocation is less than 1/10 that of sound. Though neither of these arguments is conclusive, they do attach special importance to the recognition of processes whereby a dislocation can produce a large amount of slip and can multiply without first acquiring a large kinetic energy.

We shall first show by purely topological reasoning how an unlimited amount of slip could result from the motion of a single dislocation line *ABC*, Fig. 1. For simplicity we assume that the horizontal planes are the only active slip planes. The segment *AB* is therefore fixed. A small shear stress applied on the slip plane and in the slip direction will cause the line *BC* to sweep around like the hand of a clock producing slip of one atomic spacing per revolution.

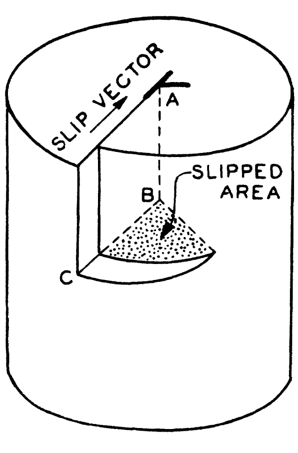

FIG. 1. Slip resulting from the motion of a single dislocation line.

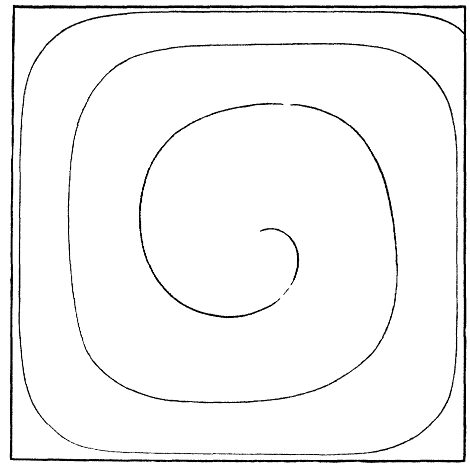

FIG. 2. Spiral form resulting from a square boundary at which slip is prevented.

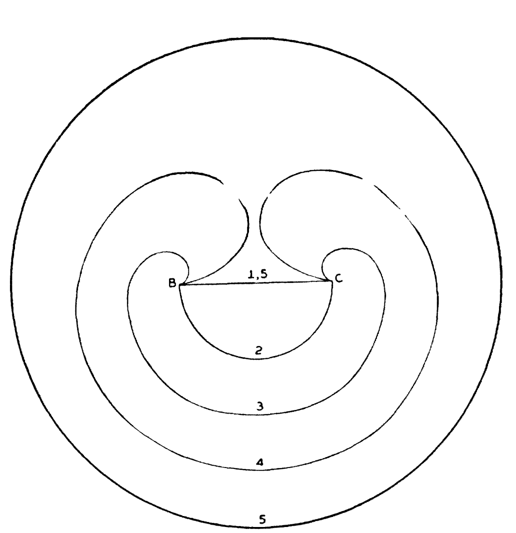

FIG. 3. Generation of successive closed loops of dislocation line.

Actually the line *BC* would not remain radial but would develop into a rotating spiral, owing to the higher angular velocity of the innermost portion. The quantitative treatment of the problem is strictly analogous to the theory of crystal growth[3] except that, when the spiral has many turns, a correction must be made for the mutual repulsion between successive turns. If slip is prevented at the boundary (as could occur in a polycrystal), the spiral would reach an equilibrium state, as is illustrated in Fig. 2 for the case of a square boundary.

Another process closely analogous to crystal growth and leading not only to continued slip but also to the generation of successive closed loops of dislocation line is illustrated in Fig. 3. The segment *BC* of a dislocation line lies in the active slip plane, the other parts of the line lying outside of the plane so that the points *B* and *C* are fixed. A suitable applied shear stress will cause *BC* to curve as shown and to generate dislocation loops at essentially the same rate as turns of the spiral were generated in the previous case. The minimum stress at which this will occur is determined by the distance *BC* and is approximately the rigidity modulus divided by the distance *BC* in lattice spacings. At a smaller stress some thermal activation will be required.

It remains to show why dislocations should have portions lying in particular glide planes and pass out of these glide planes at particular points. This is to be expected if dislocation lines have lower energy in particular crystallographic planes. A theory of the reduction in energy can be given for the particular cases of the close packed planes in h.c.p. and f.c.c. metals in terms of dissociation into "extended dislocations."[4] These metals are the ones for which slip on particular planes is most characteristic.

It is necessary for the mechanism we have described that the end points B and C should be anchored at least in some degree. This anchoring may be provided in various ways in dislocation networks, the simplest example being that in which BC is one side of a rectangular dislocation loop $ABCD$, the Burgers vector being normal to the plane of the loop. The stress which causes the motion of BC in Fig. 3 produces an opposite motion of AD, with no net force on AB and CD. In many cases (e.g., probably the hexagonal metals) the lower mobility of dislocations in planes of less than closest packing may suffice to anchor the points B and C.

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Prismatic Dislocations and Prismatic Punching in Crystals

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SMAKULA and Klein[1] have made indentation studies of the plastic properties of the thallous halides with use of a small conical punch and have found that the strain may be transmitted in a highly concentrated manner through distances large compared to the dimensions of the hole made by the punch. The system behaves as if the punch pushed cylinders or prisms of the material as rigid units in the direction of the axis of the prism, the length of the prism being relatively large compared with the sectional dimensions. The prisms appear to be polygonal in cross section, the planar surfaces being slip planes (110) and the direction of the axis being slip directions (100). A single prism having square cross section is pushed if the indenter is pressed in the (100) direction, but two and three are pushed, respectively, if the load is applied in the (110) and (111) directions. In the latter two cases the axes of the prisms lie in different, symmetrically equivalent, (100) directions. We shall term this *prismatic punching*.

In order to explain these rather dramatic results it is only necessary to discover a mechanism whereby dislocation rings of the type shown in Fig. 1 may be generated in such a way that their contours coincides with the cross section of the prismatical punchings and the Burgers vector is along the axis of the prism. If a dislocation ring of this type moves parallel to the axis of the prism, the inside of the prism will be displaced relative to the outside along the boundary by one unit of slip. Moreover, dislocations of this type are constrained to move on the surface of the cylinder or prism since their projection normal to the axis of the prism must remain unchanged if the temperature is sufficiently low that diffusion cannot occur. For this reason we shall call them *prismatic dislocations*.

It is not difficult to see how the required prismatic dislocations may be generated. The slip planes bounding the surface of the prism, which is displaced by the indenter, intersect the surface of the specimen in the region of maximal shearing stress at the periphery of the area of contact between the indenter and crystal. Thus each such bounding plane becomes the seat of spirals or rings of dislocation which can be generated in a slip plane in the manner suggested by Frank and Read in the accompanying letter. The Burgers vector associated with the rings will be in the direction of the axis of the prism. Two rings (or spirals) which are generated on different planes that intersect on a line parallel to the axis of the prism can meet at the boundary line and interact to form segments of prismatic dislocations. Complete prismatic dislocations can be formed by combining rings from each of the bounding surfaces. An example s ishown schematically in Fig. 1. The rectangular prism $ABCDEFGH$ is pushed on the area $ABFE$, which is part of the surface of the specimen, coinciding with the area of contact of the indenter. This prism is bounded by four slip planes: $ABCD$, $ADEH$, $EFGH$- and $BCGF$. The dislocation rings 1, 2, 3, and 4 shown in Fig. 1a are generated on each of these four planes. As a result of the applied shearing stress, they expand within the four planes and may extend beyond the bounding surface of the prism, as shown in Fig. 1b. The rings on different intersecting faces, such as 1 and 2 on $ABCD$ and $ADHE$, respectively, may meet at points such as a and b in Fig. 1b. The cohesion of the lines is weak at junction points such as a and b and may break under the applied stress to form two prismatic dislocations α and β in Fig. 1c and dislocation rings such as s, t, u, and v in the same diagram. The last-named rings are bent so as to lie in two slip planes which meet at the edges of the prism. The prismatic dislocations α and β have opposite signs in the sense that α is the equivalent of an extra layer of atoms equal in thickness to the Burgers vector lying in a cross section of the prism, whereas β is the equivalent of a deficiency of one plane of atoms. The first dislocation may wander down the prism to the opposite face, where it will produce a jutting by one Burgers distance over the area ($CDHG$) representing the intersection of the prism with this surface. (In practice this surface could be much farther from $ABEF$, on a relative scale, than is shown.) On the other hand the prismatic dislocation β will emerge at the surface $ABEF$ where the force is applied and produce a depression of one Burgers distance. Bent dislocations of the type s, t, u, and v may produce localized plastic flow in the vicinity of the indenter. It may be noted that these dislocations would not be generated if the rings shown in Fig. 1a did not expand beyond the boundary of the prism, as illustrated in Fig. 1b, but met tangentially at the boundary lines.

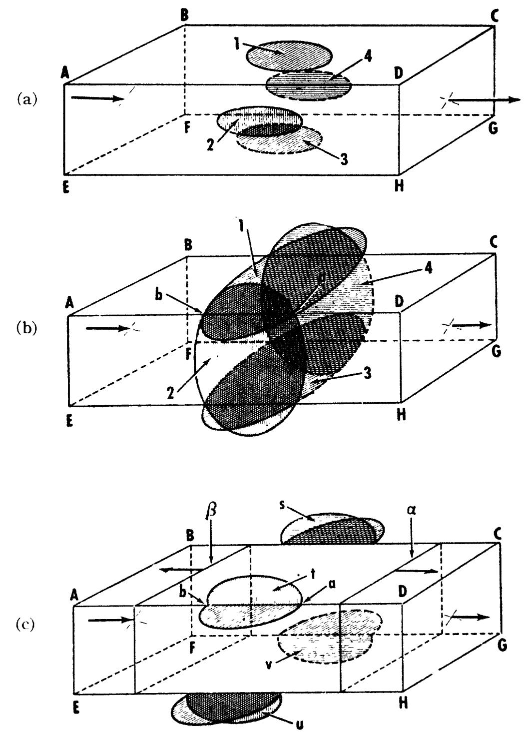

FIG. 1. Possible forms of dislocation rings.